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Procedia Engineering 2 (2010) 1731–1739

Procedia Engineering

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Fatigue 2010

Influence of thermal and thermomechanical treatments on the fatigue limit of a bainitic high-strength bearing steel

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Received 28 February 2010; revised 9 March 2010; accepted 15 March 2010

Abstract

To increase the fatigue limit of high-strength steels which typically fail from non-metallic inclusions it is necessary to modify the inclusions and/or the surrounding matrix. The goal must be a higher threshold for crack initiation and/or crack propagation. One possibility to reach this goal is to execute a thermal or thermomechanical treatment (TT and TMT). The present work discusses a series of such treatments that were applied to classical bainitic materials states.
The different treatments consisted of shortened isothermal bainitic transformations as well as cyclic loading, either with constant or stepwise increasing load amplitude, which were applied at increased temperature in the range of dynamic strain ageing. The lifetime behaviour of the different material states was determined in bending fatigue tests.
In order to describe and compare the different treatments it was necessary to evaluate the stress intensity factors arising at the critical inclusions. The results for cyclic loading yielding a lifetime of 10^7 cycles turned out to be in good agreement with the predictions using the stress intensity factor developed by Murakami. To consider different numbers of cycles to failure this calculation method was successfully extended and a modified stress intensity factor was defined. The same method was also used to test and evaluate specimens with artificial flaws. These showed the same behaviour as the specimens which failed from natural non-metallic inclusions.

Keywords: fatigue limit; high-strength steels; bainite; dynamic strain ageing; artificial flaws; non-metallic inclusions

1. Introduction

High strength steels are needed by various industries, for example the automotive industry, for the manufacture of high-end equipment components. Besides the high quasistatic requirements, such components have to withstand cyclic loadings for very long durations. Indeed, the endurance limit of these steels is at least as important as their quasistatic strength. Mostly martensitic and bainitic materials states are used for such applications.

We investigated whether and how the endurance limit of a bainitic bearing steel can intrinsically be increased by modified heat treatments. Therefore, we conducted thermomechanical treatments as well as modified isothermal bainitic transformations to optimise the bainitic microstructure to increase the endurance limit.

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1877-7058
doi:10.1016/j.proeng.2010.03.186

The investigation starts from a bainitic state of the bearing steel 100 Cr 6 (SAE 52100) which results from a conventional isothermal transformation. The thermomechanical treatment was carried out in the temperature range of maximal dynamic strain ageing following [1]. The idea was to modify the surroundings of non-metallic inclusions which are the typical crack initiations sites in high strength steels [2]. Additionally, a wide range of isothermal bainitic transformation temperatures were conducted and the resulting microstructures evaluated by tensile tests and micrographs. The goal was to find an excellent homogeneous microstructure state with endurance limits comparable with or better than our reference. The last experiment was a combination of short isothermal bainitic transformation with additional annealing. This combination led to the highest endurance limits.

2. Experimental setup

2.1. Materials states

The investigations are all carried out with the bearing steel 100 Cr 6 (SAE 52100), which is the steel most used for high strength applications. In the initial state it has a ferritic matrix with globular carbides, which have a diameter less than 1 µm.

The reference for our investigations was an isothermal transformed bainitic microstructure. The treatment was carried out in two salt baths: the first bath was used for the austenitisation at a temperature of 855 °C for 20 min. After that the specimens were transferred quickly to the second bath, which was being kept at a constant temperature of 220 °C. After 6 hours of isothermal transformation the specimens were taken out of the bath and cooled down to room temperature at ambient conditions. The resulting material state is named B220.

To increase the fatigue limit the thermomechanical treatments (TMT) were carried out with specimens in the state B220. The temperature of maximal dynamic strain ageing for our parameters was found at 275 °C. At this temperature a cyclic loading with constant amplitude of a sine wave form and a frequency of 1 Hz was chosen. Three different load histories were realised. The first was with constant amplitude of 1700 MPa applied for 10 cycles. The resulting state was named B220TMT1. The second TMT was a stepwise increase of load amplitude to the same end level as in the first TMT. The load amplitude was increased in 10 steps each with 5 cycles. The first nine steps had equidistant steps of 180 MPa, the last step had only a height of 80 MPa. The resulting state from the second TMT was named B220TMT2. The third TMT consisted of the last three steps of B220TMT2. Consequently, the resulting state was named B220TMT3. Further details of the experimental setup are described in [2].

To assert whether B220 is indeed the best bainitic state with a view to highest fatigue limits the temperature of isothermal treatment was varied between 180 °C and 300 °C in steps of 20 °C. The most promising state resulted from a transformation at 260 °C. For fatigue tests this state named B260 was realized with an isothermal transformation time of only 2 hours because measurements of the retained austenite showed that the transformation of austenite was completed after that time.

The last bainitic modification was an isothermal transformation like to reach the state B220, but it was stopped after just one hour of isothermal transformation at 220 °C. After cooling down to room temperature the specimens were annealed at 180 °C for 2 hours, as it is usual for martensitic states with highest fatigue limits. This state was named B220A

2.2. Specimens geometry

All specimens were machined with grinding allowance before the heat treatment and were grinded to final shape afterwards.

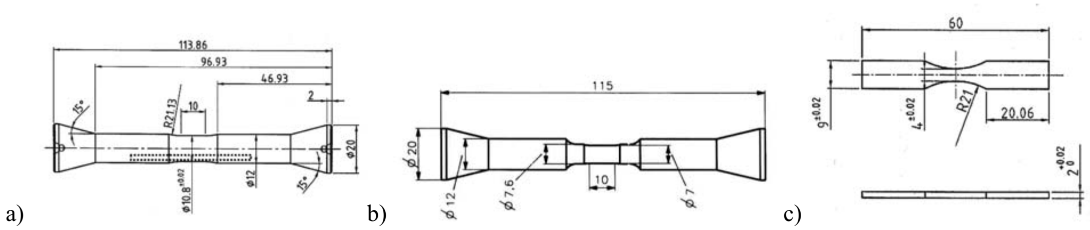


Fig. 1. (a) Specimens for thermomechanical treatment, (b) tensile testing, and (c) bending fatigue tests

The thermomechanical treatment was applied to the specimens given in Fig 1a. Out of them either one tensile test specimen (Fig 1b) or three flat fatigue specimens (Fig 1c, one of the three positions of a fatigue specimens is given with dashed lines in Fig 1a) were machined to their final shape so that the areas of the gauge length of all of these specimens were the same.

2.3. Testing conditions

The microstructure was investigated in cross sections, which were grinded, polished and etched with Nital. The amount of retained austenite was measured by X-ray diffraction [3].

The tensile tests were carried out with standard tensile test equipment at room temperature. Deformation speed was 10^{-4} s^{-1}. The strain was measured in the gauge length with strain gauges.

Hardness is given in Vickers as the average of at least 10 single measurements.

Fatigue tests are set up with alternating bending at a frequency of 25 Hz. The ultimate number of cycles was 10^7. The endurance limit was estimated by using the staircase method [4].

2.4. Evaluation of fracture surfaces

The distribution of the size and of the location of the non-metallic inclusions causes a significant part of the scatter of the number of cycles to failure, as displayed in the results. Therefore, for the purpose of analyzing the location and the morphology of each inclusion at which crack initiation has taken place it was necessary to calculate the individual stress intensity factor for each location individually. The maximum stress intensity factor K_{max} acting at an inclusion is given by:

$$K_{\max} = B \cdot \sigma_{\max} \sqrt{\pi \sqrt{A}} \qquad (1)$$

with B = 0,65 for inclusions at the surface and B = 0,5 for inclusions below the surface [1]. σ_{max} is the maximum normal stress at the critical position of the inclusion where crack initiation takes place. This is the position nearest to the surface for inclusions that have no contact to the surface, and the opposite side for inclusions which are located at the surface. A is an elliptical area enclosing the inclusion.

To understand the results it was necessary to consider the decreasing hardness, which is a consequence of the annealing during TMT and which is supposed to reduce the stress intensity factor. The question was addressed by applying the following formula proposed by Murakami [5]:

$$K_{th} = C(HV + 120)\sqrt{A}^{\frac{1}{3}} \tag{2}$$

with C depending on the position of the inclusion and the Vickers hardness HV. Following [2] this equation may be modified by the number of cycles to failure N_f:

$$K_{\text{mod}} = (D - E \cdot \log N_f)(HV + 120)\sqrt{A}^{\frac{1}{3}}. \tag{3}$$

with $D = 5.05 \cdot 10^{-3}$ for inclusions at the surface, $D = 4.7 \cdot 10^{-3}$ for inclusions below the surface and $E = 0.27 \cdot 10^{-3}$ for both types of inclusions. The ratio between K_{max} and K_{mod} is plotted versus the number of cycles to failure and gives the degree of change of the stress intensity factor caused by the TMT and TT: a ratio of 1 means that there is no change, a ratio higher than 1 shows that the stress intensity factor K_{max} is higher than expected by Equation 3.

2.5. Artificial flaws

To simulate the natural non-metallic inclusions artificial flaws were positioned in the middle of the testing surface of the fatigue specimens by die sinking. The size of the flaws was varied by using different diameters of the die sinking electrode so that we attained die sinked flaws and micro die sinked flaws [6]. The so prepared specimens were tested and evaluated by the methods described above.

3. Results

3.1. Thermomechanical treatments

The cyclic deformation behaviour during the thermomechanical treatment at 275 °C is given in Fig 2. The curve of the TMT1 shows cyclic hardening. The amplitude was derived from a few tests with different amplitudes with the aim to have a high plastic strain in the beginning of the treatment followed by cyclic hardening. Higher amplitudes led to cyclic softening after a first hardening (see curves for $\sigma_a > 1700$ MPa in Fig 2). TMT2 with stepwise increased amplitudes shows a neutral behaviour in the first steps and a very slight hardening in the following steps. Obviously the behaviour during the last three steps of TMT2 was significantly influenced by the preceding steps, because TMT3, whose steps were identical to the last three steps of TMT2, shows higher plastic deformation and strong hardening in the first step followed by slight hardening in the second and neutral behaviour in the third step. The behaviour of TMT3 is more similar to TMT1 than to TMT2.

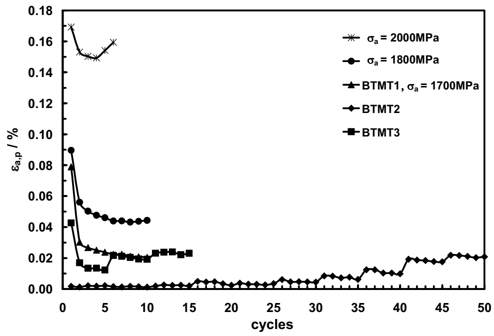


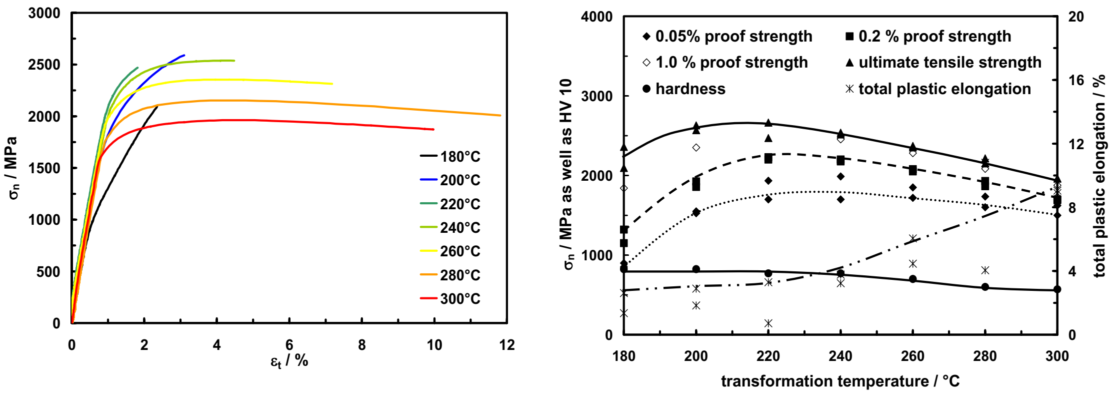
Fig. 2. Cyclic deformation curves of the different thermomechanical treatments

3.1.1. Tensile tests

Fig 3 shows the results of tensile and hardness tests in material state B after the various heat treatments with the different isothermal transformation temperatures.

Individual tensile test curves show the highest strength between 200 °C and 240 °C but the total elongation to fracture is higher for higher transformation temperatures (Fig. 3a). Remarkable is the tensile behaviour after transformation at 180 °C: there is a change in the slope of the tensile test curve at about 850 MPa, which probably indicates that the isothermal transformation led to a material state with retained austenite which is not stable during tensile tests and transforms stress or strain induced during tensile tests into martensite.

Fig 3 b gives an overview based on two tensile tests for each transformation temperature. Each symbol marks an individual result while the lines symbolize the trends. The hardness continuously decreases with increasing transformation temperature. In contrast, the ultimate tensile strength has its maximum at about 220° C. The yield strength appears to reach its maximum with a transformation temperature between 220° and 240° C. The total plastic elongation to fracture is quite low for low transformation temperatures but increases significantly at transformation temperatures between 240 °C and 300 °C.



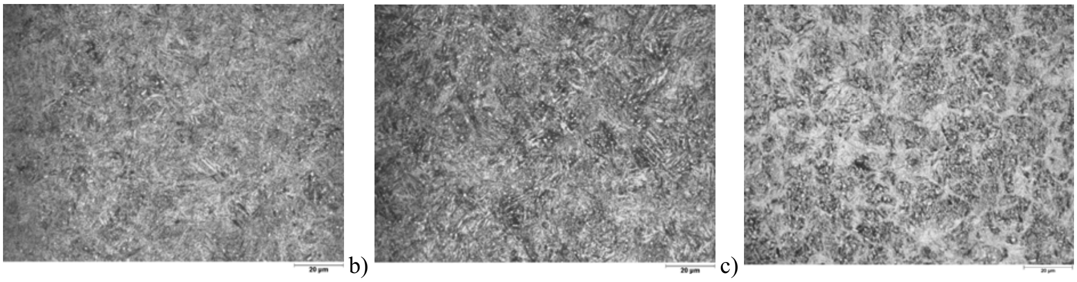
a) b)

Fig. 3. Results of the tensile and hardness tests (a) individual curves; (b) Overview and trends of all tests

3.2. Micrographs and hardness

The microstructure is mainly bainitic for all transformation temperatures, although at a transformation temperature of 180° C there is also a low amount of martensite. Neither microscopic nor roentgenographic examinations revealed any retained austenite, except in the state B220A, which contained an amount of retained austenite of 7.5 volume percent. The microstructure of the states B220, B260, and B220A are given in Fig 4. In these figures as well as in all other states the structural dimensions of the microstructure increase with increasing transformation temperature. Even a transformation temperature of 300° C leaves only lower bainite in the microstructure. All states show a good bainitic microstructure with an appropriate distribution of carbides with diameters of less than 1 μm.

The hardness of B220 is 770 HV 10. The TMT decreases the hardness to 725 HV 10 in all three cases. B260 has a hardness of 700 HV 10 and B220A reaches 725 HV 10.



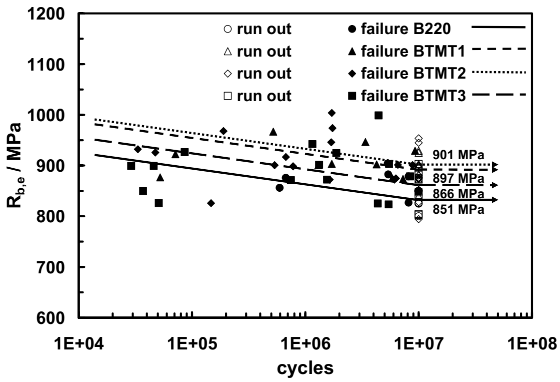
a) b) c)

Fig. 4. Micrographs of etched cross sections of the states (a) B220; (b) B260, and (c) B220A

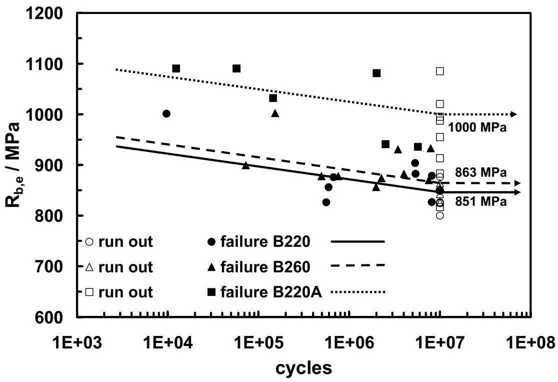
3.3. Fatigue tests

Fig 5 summarizes the results of the fatigue tests. Fig 5a shows the influence of the TMT on the fatigue limit. Although the TMT reduces the hardness of the specimens, the fatigue limit is increased in all three cases as compared to the state B220. The maximum increase is nearly 6 %. TMT2 led to the highest fatigue limit while TMT3 increases the fatigue limit just slightly. It is obvious that the scatter in the number of cycles to failure is huge, which makes it difficult to find out a clear trend between the different TMT.

While the scatter in Fig 5b is still large there is a clearly visible gap between the two completely isothermally transformed states B220 and B260 and the state B220A. The shortened isothermal transformation with the additional annealing of B220A results in an increase of the fatigue limit of 17.5 % as compared to B220. In contrast, B260 has just a slightly higher fatigue limit than B220. This is interesting also because the hardness and the tensile strength of B260 are significantly lower compared to B220.



a)

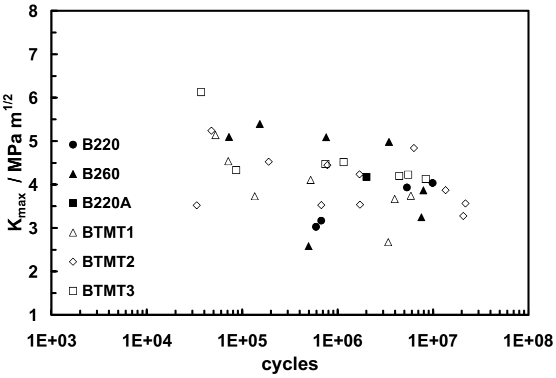


b)

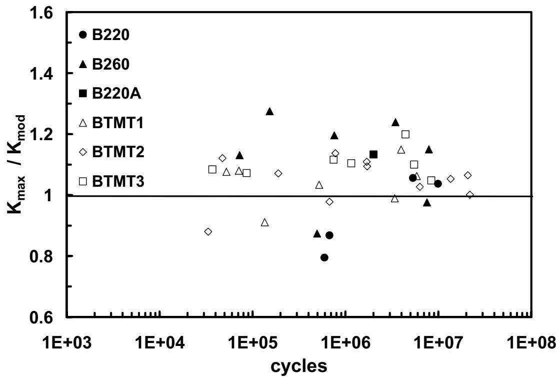
Fig. 5. SN-curve of (a) B220 and the TMT; (b) B220, B260, and B220A

3.4. Evaluation of the fracture surfaces

The wide scatter in the number of cycles to failure is due to the variation in the crack initiation site, which in most cases is a non-metallic inclusion. The remaining specimens failed from initiation sites at the surface without signs of non-metallic inclusions. To consider the influence of the size and the location of the critical inclusion on the fatigue limit the fracture surfaces were analysed in the way described above (see 2.4). The plotting of K_{max} versus number of cycles to failure is given in Fig 6a. As expected, a decreasing stress intensity factor results in an increasing lifetime. Surprisingly, at first sight the different material states do not show different stress intensity factors. Fig 6b plots the ratio between K_{max} and K_{mod} versus the number of cycles to failure and gives the degree of change of the stress intensity factor caused by the TT and the TMT. Table 1 contains the mean values of each state. The classical TT, which leads to B220 has a ratio smaller than 1. The TMT increases the ratio significantly to values greater than 1. The temperature treatments had still higher ratios than the TMT although there is only one fracture surface which can be analysed in the state B220A.



a)



b)

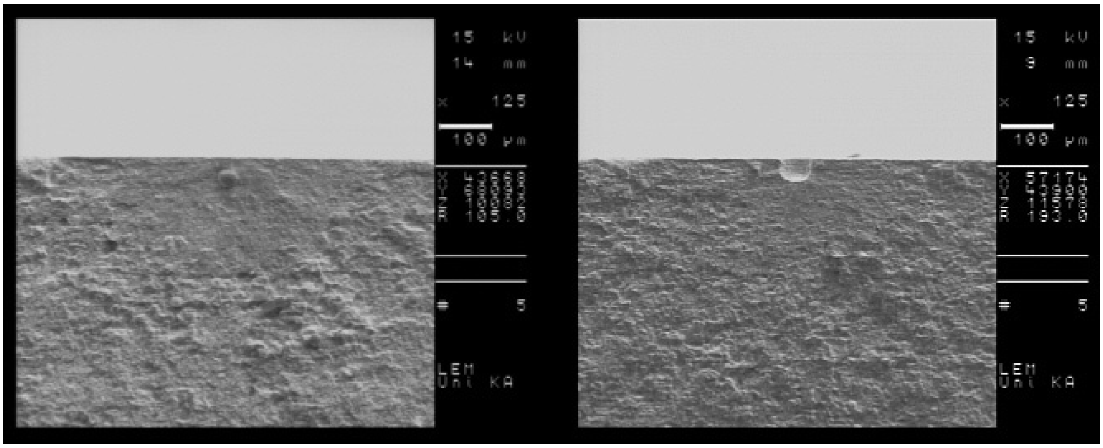
Fig. 6. (a) K_{max} and (b) ratio between K_{max} and K_{mod} versus number of cycles

Table 1. Average ratio of K_{max} / K_{mod} for the different material states

materials state	Average of K_{max} / K_{mod}
B220	0.94
B220TMT1	1.05
B220TMT1	1.08
B220TMT1	1.10
B260	1.12
B220A	1.13

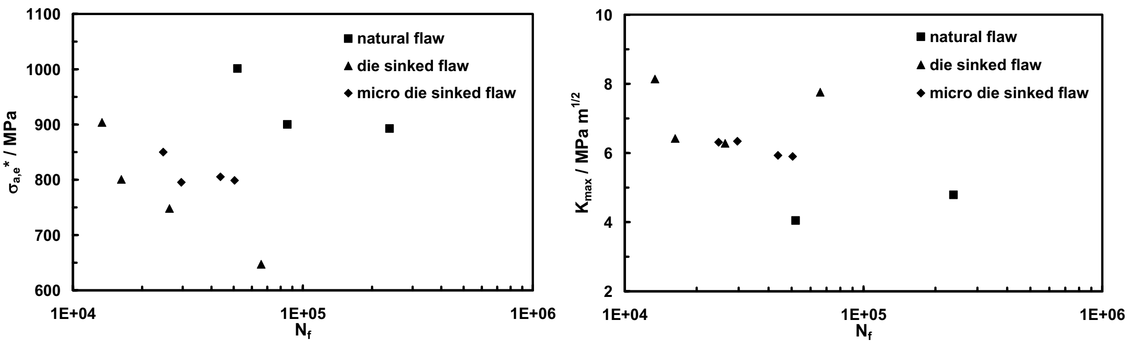
3.5. Artificial flaws

The evaluation of the fatigue limit of different microstructures is difficult in high strength steels because of the above presented scatter in crack initiation sites. A comparison of different states would be easier if the flaw sizes and locations of the individual specimens were all the same. This could be reached when crack initiation takes place at artificial flaws. To find out whether artificial flaws behave in the same way as natural flaws adequately prepared specimens were tested. Fig 7 compares a fracture surface of a specimen which failed from a natural flaw with another which failed from an artificial flaw. The crack surface morphology is the same for both types of flaws. But the sizes of the two types of flaws are different, which results in different lifetimes. Specimen with bigger artificial flaws typically failed at shorter lifetimes, which is shown in the SN-plot (Fig 8a). Since micro die sinking can produce smaller artificial flaws than normal die sinking, the micro die sinked flaws have a longer lifetime. The evaluation of stress intensity factors brings nearly all the flaw populations to one line in the K_{max}-N_f-plot (Fig 8b).



a) b)

Fig. 7. Typical fracture surface of crack initiation from (a) natural and (b) artificial flaws



a) b)

Fig. 8. (a) SN-curve for different flaw types and (b) K_{max} versus number of cycles to failure

4. Discussion

The TMT may cause an increase of the fatigue limit due to a stabilisation of the dislocation structure. The TMT in the temperature range of dynamic strain ageing has two main effects: first the dislocation density is increased by plastic deformation in the first cycle of TMT1. Second the mobile dislocations get trapped by diffusing carbon atoms, which prefer to diffuse to the core of the mobile dislocations. The consequence is the observed cyclic hardening, which could be seen for TMT1 in Fig 2. The idea for TMT with stepwise increasing load amplitudes was to achieve an advantageous modification of the surroundings of the non-metallic inclusions by such a treatment. The stress concentration at inclusions causes plastic deformation just around the inclusions at relatively low imposed stresses. Then the dynamic strain aging process may have its strongest effect of stabilisation of the dislocation structure just around the non-metallic inclusions, which are the preferred crack initiation sites. The comparison of the three TMT does not permit to take a clear decision which variant of the TMT is the best, because the fatigue limits, as well as the ratios of the stress intensity factors, all result in quite similar values. The results are clearer for the martensitic material states, where the stepwise increased TMT resulted in a significantly higher fatigue limit [2]. Nevertheless, TMT in the temperature range of dynamic strain ageing is a possibility to increase the fatigue limit also of bainitic steels, but such treatments are very costly

Therefore, we tried out two other possibilities. The idea for the first was prompted by the fact that in high strength steels the combination of high quasistatic strength and ductility is beneficial for a high fatigue limit. With the help of tensile tests we could identify an isothermal transformation temperature of 260 °C as an interesting trade off between strength and ductility. Although the resulting state showed lower hardness the fatigue limit is higher than that of B220. Hence, the manufacturing process is reduced of 4 hours heat treatment because the transformation to bainite lasts only two hours at 260 °C in contrast to six hours at 220 °C.

The second attempt to increase the fatigue limit by modified temperature treatment is a short isothermal transformation with additional annealing. This led to very good fatigue behaviour. Apart from the fine microstructure the amount of retained austenite is important for this behaviour because the austenite can reduce stress concentrations by stress or strain induced transformation of the retained austenite into martensite. This transformation is combined with a local increase of residual compression stresses, which hinder crack initiation and crack propagation.

Artificial flaws realised by die sinking have the same impact on the crack initiation process as natural flaws. So it is possible to compare different high strength steel states by testing specimens with artificial flaws. The advantage of the artificial flaws is that the scatter in lifetime, which is typically relatively big if crack initiation takes place at non-metallic inclusions with stochastic distributed sizes and locations, is drastically reduced. So the number of tests can be reduced by using specimens with artificial flaws.

5. Conclusions

The increase of the fatigue limit of high-strength steels with bainitic microstructure which fail from non-metallic inclusions is possible by a modification of the microstructure especially at the surrounding matrix of the inclusions by thermal or thermomechanical treatments. The reference for our investigations was a bainitic microstructure state resulting from an isothermal transformation at a temperature of 220 °C for 6 hours. The main conclusions are:

- A thermomechanical treatment in the temperature range of maximal dynamic strain ageing increases the fatigue limit and the ratio K_{max}/K_{mod}.
- The bainitic state B260 exhibits good fatigue resistance even when the material state has a lower ultimate tensile strength but a higher ductility. The advantage of this state is a shortened isothermal transformation time because the isothermal transformation takes place at higher temperatures.
- Without lowering the fatigue limit the transformation time can also be reduced by annealing the specimens after the isothermal transformation. The resulting state showed the highest fatigue limit in our investigation.
- Artificial flaws are a possibility to benchmark the influence of microstructural changes on the fatigue behaviour with a relatively small number of specimens in a relatively short time.

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